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⑥ THE STRENGTHENING MECHANISM

OF

② NA FERROUS MARTENSITE, ③ NA

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⑩ By

⑦ NA G. S. Ansell and A. Arrot*

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* Ford Scientific Laboratory, Dearborn, Michigan.

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Abstract

↙ A model is proposed to account for the observed strengthening behavior of ferrous martensites. The model is based upon the inheritance of carbon rich regions by the martensite which were present in the austenite prior to transformation. These carbon rich regions strengthen the martensite in a manner analogous to the strengthening effect due to a finely dispersed second phase. ↗

Introduction

Ferrous martensite is formed by a nucleation and shear mechanism when austenite is cooled, at a rate which is fast enough to suppress the formation of pearlite, below a critical temperature (M_s). Although the mechanism of the transformation process appears to be well understood,^(1,2) the mechanism responsible for the observed high strengths exhibited by ferrous martensites is uncertain.

In this paper, a model is proposed to account for the strengthening of ferrous martensite.

Model

The maximum solubility of carbon in austenite is 1.98 wt. per cent as compared with a maximum solubility of 0.025 wt. per cent in ferrite. Prior to transformation, the carbon in the austenite is nonuniformly distributed. In addition to the statistical type of composition fluctuations and undissolved carbides, the carbon segregates at dislocations, sub-boundaries, twin boundaries, grain boundaries, and other structural defects in the austenite. As will be shown later, this segregation results in regions in the austenite, particularly around dislocations, that have many times the nominal carbon composition of the steel. As the austenite is cooled, the equilibrium composition of these carbon-rich areas increases. When the austenite is cooled to the M_s temperature and below, the nucleation and shear transformation occurs. As the transformation interface moves through the structure, the defect structure present in the austenite is destroyed. The transformation is,

however, so rapid that the carbon which has segregated about these defects cannot diffuse away, and is inherited by the martensite. Indeed, one would more reasonably expect the formation of a carbide rather than the disappearance of these carbon-rich areas.

The martensite has, therefore, regions which contain many times greater carbon content than the nominal carbon composition. Since these regions exist in what already is a super-saturated solid solution of carbon in iron, these areas are either highly strained, or form carbides. Due either to the high localized strain or to the carbide formation, the elastic constants of these areas should increase as compared to the remainder of the martensite. If one now considers the movement of dislocations through the martensite, the effect of these regions upon dislocation motion should be similar to that due to a finely dispersed second phase.

Martensite may be considered, on this basis, as an alloy containing finely dispersed regions which act in a manner similar to that of a dispersed second phase. The geometry of this dispersion is dictated by the nature and degree of carbon segregation about defects, and by the distribution of the defects in the austenite prior to transformation.

Calculations

1) Segregation of Carbon in Austenite

It is assumed that the major cause of localized segregation of carbon in austenite is due to atmosphere formation about dislocations. (3)
Adopting the method of Cottrell and Bilby, the degree and character of the segregation of carbon around edge type dislocations in

austenite due to hydrostatic interactions is calculated.

The hydrostatic interaction energy, V , between the carbon atoms and edge dislocations is given by:

$$V = \Delta v \cdot \frac{\mu b}{3\pi} \cdot \frac{1+\nu}{1-\nu} \cdot \sin \frac{\alpha}{r} \quad \text{for } r \geq r_0$$

$$V = 0 \quad \text{for } r < r_0$$
(1)

where $\Delta v = 1.55 \times 10^{-24} \text{ cm}^3$, volume change due to the addition of one carbon atom to the lattice.

$\mu = 8.8 \times 10^{10} \text{ dynes/cm}^2$, shear modulus of austenite

$b = 2.58 \text{ \AA}$, Burger's vector of the dislocation

$\nu = 0.3$, Poisson's ratio

α and r are the position coordinates relative to the dislocation line, α being measured from the slip direction.

r_0 = dislocation core radius.

The atmosphere segregation of carbon about the dislocation, therefore, follows the distribution function:

$$C(r, \alpha) = C_0 \exp(-V(r, \alpha)/kT) \quad (2)$$

where $C(r, \alpha)$ is the carbon concentration at position r, α , and C_0 is the equilibrium concentration of carbon in the lattice, k is Boltzmann's constant, and T is the absolute temperature.

Using the values, $\alpha = \frac{3\pi}{2}$, since the maximum interaction occurs directly below the dislocation line and $T = M_s$ temperature in equations (1) and (2),

the ratio of C/C_0 as a function of r directly below the dislocation line was calculated and is shown in Fig. 1. for several typical values of M_s .

2) Volume Fraction of Carbon Rich Regions.

The volume fraction of carbon rich regions in the martensite is the product of the length of dislocation lines per cm^3 in austenite or the dislocation density, ρ , times the cross sectional area of

the carbon atmosphere about the dislocation lines where the carbon content is significant. The cross sectional area of the atmosphere is calculated as the semicircular area under the edge dislocation whose radius is given by the distance below the dislocation where the carbon concentration is at least 2.5 wt. per cent, or 10 at. per cent carbon. This radius is given by solution of equation (2) for a particular nominal carbon concentration in the steel. In Fig. 1, a line is shown which gives the C/C_0 value required to produce 2.5 wt. per cent carbon as a function of the nominal carbon content of the steel. One may therefore determine from Fig. 1 the radius of the atmosphere knowing the nominal carbon content of the steel and its M_s temperature assuming that the atmosphere reaches equilibrium concentration prior to transformation.

The fraction of the original solute, F , which has segregated to the dislocations in any time, t , follows the equation (4)

$$F = 1 - \exp \left[-3 \rho \left(\Delta v \cdot \frac{ub}{3\pi} \cdot \frac{1+\nu}{1-\nu} \cdot D \cdot t / AT \right)^{2/3} \right] \quad (3)$$

where D is the diffusion rate of carbon in austenite. From this equation it is obvious that in the case of plain carbon and low alloy steels where the M_s is sufficiently high, the equilibrium atmosphere forms completely almost instantaneously about the dislocations (1 sec. at 500°C), and is present prior to transformation. For those alloys where the M_s temperature is reduced to room temperature and below, the time to form a complete equilibrium atmosphere increases, and either some evidence of a quenching rate

effect upon the atmosphere or the formation of atmospheres characteristic of some higher temperature might be found.

The volume fraction, f , of the retained carbon atmosphere in martensite is therefore given by the relation:

$$f = \frac{\pi r_a^2}{2} \cdot \rho \quad (4)$$

where r_a is the atmosphere radius and ρ is the austenitic dislocation density.

3) Yield Strength of Martensite

(5)

Ansell has derived on the basis of dislocation theory that, for resolved stresses, the yield strength, $\sigma_{y.s.}$, of an alloy containing a dispersion of second-phase spherical particles of less than 800Å diameter is given by the relation:

$$\sigma_{y.s.} = \frac{\mu^*}{120} \cdot f^{1/3} \cdot \frac{1}{(0.82 - f^{1/3})} + \sigma_{wh.} \quad (5)$$

where μ^* is the shear modulus of the dispersed phase and $\sigma_{wh.}$ is the increment of stress due to work hardening between the strain associated with first yielding and the strain of the measured yield stress.

When the volume fraction, f , of the dispersed phase is very small, as in the case of martensite, this equation may be rewritten:

$$\sigma_{y.s.} = \frac{\mu^*}{98} \cdot f^{1/3} + \sigma_{wh.} \quad (6)$$

where μ^* is approximately 3×10^{12} dynes/cm².

For unresolved stresses the yield strength of martensite is therefore given by the equation:

$$\sigma_{y.s.} = 10^{11} \text{ dynes/cm}^2 (1.57 r_a^2 \rho)^{1/3} + \sigma_{wh.} \quad (7)$$

Discussion and Conclusions.

The calculations based upon the proposed model predict essentially the same strength dependence on carbon content as has been predicted by other models. (6) The model, however, additionally predicts that the strength of ferrous martensites is sensitive to both the defect structure of the austenite prior to transformation and the temperature at which the austenite transforms to martensite.

It is the effect of the prior austenitic substructure on the properties of ferrous martensite which makes this model unique. The model, therefore, readily explains the effects observed in the use of the ausform process in the production of a martensitic structure with increased strength and ductility as compared with the properties of martensite produced by the normal quench and temper process. (7)

Although it cannot be demonstrated that the strength of ferrous martensites is uniquely explained in this manner, this approach should be considered as one alternative mechanism which could account for the observed strengthening behavior of ferrous martensite.

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FIGURE CAPTIONS

Figure 1. The ratio of carbon segregation as a function of distance directly below a dislocation line is shown for several typical values of M_s . A line is drawn to indicate the ratio of segregation required to produce 2.5 wt. per cent carbon segregate as a function of nominal carbon content.

